Interim Technical Report

from

Department of Metallurgical Engineering
and Materials Science
Carnegie Mellon University

to

Air Force Office of Scientific Research
Electronic and Solid State Sciences
Bolling Air Force Base
Washington, DC 20332

on

Fundamental Studies of Beta Phase Decomposition Modes in Titanium Alloys

by

H. I. Aaronson, Principal Investigator
T. Furuhara, Graduate Student
Y. Mou, Graduate Student

for the period
1 October 1987 - 30 September 1988

31 January 1989

SECURITY CLASSIFICATION OF THIS PAGE

AD-A205 296

REPORT DOCUMENTATION PAGE

Form Approved
OMB No. 0704-0188

1a. REPORT SECURITY CLASSIFICATION UNCLASSIFIED	1b. RESTRICTIVE MARKINGS
2a. SECURITY CLASSIFICATION AUTHORITY	3. DISTRIBUTION / AVAILABILITY OF REPORT UNLIMITED
2b. DECLASSIFICATION / DOWNGRADING SCHEDULE	
4. PERFORMING ORGANIZATION REPORT NUMBER(S)	5. MONITORING ORGANIZATION REPORT NUMBER(S) AFOSR-TR- 89-0235
6a. NAME OF PERFORMING ORGANIZATION CARNEGIE-MELLON UNIVERSITY 6b. OFFICE SYMBOL (If applicable)	7a. NAME OF MONITORING ORGANIZATION AFOSR/NE
6c. ADDRESS (City, State, and ZIP Code) DEPARTMENT OF METALLURGICAL ENGINEERING AND MATERIALS SCIENCE	7b. ADDRESS (City, State, and ZIP Code) BLDG. 410 BOLLING AFB, DC 20332
8a. NAME OF FUNDING / SPONSORING ORGANIZATION AFOSR 8b. OFFICE SYMBOL (If applicable) NE	9. PROCUREMENT INSTRUMENT IDENTIFICATION NUMBER AFOSR-84-0303
8c. ADDRESS (City, State, and ZIP Code) BLDG. 410 BOLLING AFB, DC 20332	10. SOURCE OF FUNDING NUMBERS PROGRAM ELEMENT NO: 61102F; PROJECT NO: 2306; TASK NO: A1; WORK UNIT ACCESSION NO:

11. TITLE (Include Security Classification)
FUNDAMENTAL STUDIES OF BETA PHASE DECOMPOSITION MODES IN TITANIUM ALLOYS

12. PERSONAL AUTHOR(S)
H.I. Aaronson

13a. TYPE OF REPORT	13b. TIME COVERED	14. DATE OF REPORT (Year, Month, Day)	15. PAGE COUNT
ANNUAL TECHNICAL	FROM 01 Oct 87 TO 30 Sep 88	Jan 30, 1988	

16. SUPPLEMENTARY NOTATION

17. COSATI CODES			18. SUBJECT TERMS (Continue on reverse if necessary and identify by block number)
FIELD	GROUP	SUB-GROUP	

19. ABSTRACT (Continue on reverse if necessary and identify by block number)

A TEM investigation of the interphase boundary structure of (hcp) grain boundary alpha allotriomorphs in a Ti-7.15 W/O Cr alloy has shown that whether these crystals are Burgers or non-Burgers-related with respect to their bounding (bcc) beta matrix grains their interfacial structures are partially coherent. No misfit dislocations are observed at either type of interface. Instead, two types of ledge are present. One type is widely and irregularly spaced and heavily kinked; this appears to be growth ledges. The other type are quite straight and both closely and uniformly spaced; this is likely to be structural ledges. Habit plane measurements on the broad faces of alpha plates support the presence of structural ledges. These ledges have a Burgers vector of a/6 (111) parallel to the terrace plane, indicating that they can perform as significant compensators of misfit dislocations. Studies on the interphase boundary structure of the BCC/HCP amssive transformation in a Ag-26 A/O Al alloy have reached the point where a technique for preparing thin foils in a manner which permits a significant number of interfaces to be imaged has finally been achieved. Studies of this structure have now been initiated.

20. DISTRIBUTION / AVAILABILITY OF ABSTRACT	21. ABSTRACT SECURITY CLASSIFICATION
☒ UNCLASSIFIED/UNLIMITED ☐ SAME AS RPT ☐ DTIC USERS	UNCLASSIFIED
22a. NAME OF RESPONSIBLE INDIVIDUAL DR ROSENSTEIN	22b. TELEPHONE (Include Area Code) (202) 767-4933 — 22c. OFFICE SYMBOL NE

DD Form 1473, JUN 86 — Previous editions are obsolete — SECURITY CLASSIFICATION OF THIS PAGE

TABLE OF CONTENTS

Fundamental Studies of Beta Phase Decomposition Modes in Titanium Alloys

ABSTRACT

A TEM investigation of the interphase boundary structure of (hcp) grain boundary alpha allotriomorphs in a Ti-7.15 W/O Cr alloy has shown that whether these crystals are Burgers or non-Burgers-related with respect to their bounding (bcc) beta matrix grains their interfacial structures are partially coherent. No misfit dislocations are observed at either type of interface. Instead, two types of ledge are present. One type is widely and irregularly spaced and heavily kinked; this appears to be growth ledges. The other type are quite straight and both closely and uniformly spaced; this is likely to be structural ledges. Habit plane measurements on the broad faces of alpha plates support the presence of structural ledges. These ledges have a Burgers vector of a/6 <111> parallel to the terrace plane, indicating that they can perform as significant compensators of misfit in this interface and thereby eliminate or greatly reduce the driving force for acquisition of misfit dislocations. Studies on the interphase boundary structure of the (bcc) $\beta \rightarrow$ (hcp) ξ_m massive transformation in a Ag-26 A/O Al alloy have reached the point where a technique for preparing thin foils in a manner which permits a significant number of β:ξ_m interfaces to be imaged has finally been achieved. Studies of this structure have now been initiated.

1. INTRODUCTION

This program consists of inter-related fundamental studies on the crystallography, morphology and kinetics of the proeutectoid alpha and the massive alpha transformations in Ti-X alloys. Because that portion of the beta matrix which is not transformed to massive alpha is converted to martensite during quenching to room temperature in the Ti-X systems in which the $\beta \rightarrow \alpha_m$ reaction has been observed (1), however, the crystallographically equivalent massive transformation in a Ag-26 A/O Al alloy has been (2) and is currently being employed as a surrogate model massive transformation.

The experimental portions of the present program are centered about studies of interphase boundary structure. The interphase boundaries being examined are generated during the precipitation of grain boundary alpha allotriomorphs at beta grain faces in a Ti-7.15

W/O Cr alloy and during the $\beta \rightarrow \xi_m$ massive transformation in an Ag-26 A/O Al Alloy. A more theoretically oriented program, in which results obtained during the present program since its inception in 1973 are being utilized to seek fundamental generic differences between shear and diffusional mechanisms of growth, is now being developed by the P.I. This is presently the major (though by no means the only) component of the "bainite controversy" now being fought out by physical metallurgists around the world. Although originally focussed primarily upon the bainite reaction in steel, with rapid expansion of knowledge of phase transformations in non-ferrous systems (and even in ceramics), the materials base of these arguments has now been broadened to include first order phase changes in any crystalline solid in which the product phase (or phases) is also crystalline. With the support of this grant, the P.I. gave the introductory lecture at the International Conference on Bainite held during the World Materials Congress in Chicago in September, 1988 under the direction of Prof. Morris Cohen. Although this conference helped to focus attention upon the more critical aspects of the bainite controversy, the individual disputes presently comprising this controversy seemed to become even more intense, with particular interest being focussed by Asians and Europeans as well as Americans upon the shear vs. diffusional growth argument in which we are actively participating.

2. CRYSTALLOGRAPHY, INTERFACIAL STRUCTURE AND GROWTH KINETICS OF GRAIN BOUNDARY ALPHA ALLOTRIOMORPHS IN A TI-7.15 W/O Cr ALLOY

2.1 Introduction

Studies of the nucleation kinetics of grain boundary allotriomorphs of proeutectoid ferrite during the proeutectoid ferrite reaction in Fe-C (3) and in Fe-C-X (4) alloys and during the proeutectoid alpha reaction in Ti-X alloys (5) have provided indirect but nonetheless rather convincing evidence that grain boundary allotriomorphs ought to be as coherent as possible during their nucleation stage (6). On the considerations of van der Merwe (7), during the early stages of growth, most coherent interfaces--unless the misfit across them is very small indeed and/or mechanisms for acquisition of misfit dislocations are scarce or operating very slowly--should become partially coherent (6). This view contravenes earlier wisdom on grain boundary nucleated precipitates (8,9), according to which the allotriomorph:matrix interface is likely to be partially coherent with respect to only one of the matrix grains forming a grain face. A disordered structure is likely to form the interface between the allotriomorph and the adjacent

matrix grain, since the orientation relationship between the allotriomorph and the latter grain was presumed to be irrational and thus incapable of supporting a partial coherent interfacial structure. However, measurements of the thickening kinetics of ferrite allotriomorphs in Fe-C alloys, wherein the collector/rejector plate mechanism (10) is unlikely to interfere, yielded rates somewhat less than those allowed by volume diffusion control (11), suggesting that partial coherency might reign on both broad faces of grain boundary allotriomorphs. Although experimental observations indicate that ferrite allotriomorphs have a rational orientation relationship with respect to only one austenite grain (12), facets are readily observed on both the interface of the allotriomorph with this austenite grain as well as the interface with the austenite grain toward which the allotriomorphs are irrationally oriented. A similar observation was later made on grain boundary allotriomorphs of ξ_m formed during the $\beta \rightarrow \xi_m$ massive transformation in a Ag-26 A/O Al alloy (2). Hence the available indirect evidence on both nucleation and growth suggests that allotriomorphs are probably partially coherent with respect to both matrix grains forming grain faces. At the present time, however, no direct evidence is available in the literature on interfacial structure of grain boundary allotriomorphs. The present investigation was undertaken to fill this important gap in our knowledge and will serve as the Ph.D. thesis of Mr. Tadashi Furuhara.

Although it might have been preferable to have conducted this investigation on the proeutectoid ferrite reaction, since this is by far the most widely known and studied one in solid metallic alloys, conversion of effectively all of the austenite untransformed during isothermal reaction to martensite during quenching to room temperature makes observation of the interphase boundary structure of ferrite allotriomorphs in hypoeutectoid steels impossible with room temperature techniques. (Use of hot-stage TEM could circumvent this problem, but the loss of resolution attending observations made at elevated temperatures would have even more serious consequences). On the other hand, in numerous hypoeutectoid Ti-X alloy systems it is possible to retain all of the beta matrix upon quenching to room temperature, even in thin foils, once a critical composition is exceeded. We chose a hypoeutectoid Ti-7.1%W alloy for the present study because we have used the same alloy in several previous investigations and have found it quite satisfactory for the purposes of the present study. Some very small precipitates of the transitional omega phase form in the beta matrix during quenching to room temperature, but they do not seriously interfere with TEM observations.

In our previous Interim Technical Report, we have shown that grain boundary alpha allotriomorphs in a Ti - 7.15 W/O Cr alloy have an exact or near-Burgers orientation relationship (13) with respect to one of its parent β grains but an irrational one with respect to the other beta grain (14). All of the Burgers-related interfaces examined were found to contain one or more sets of growth ledges as well as one set of what are presently presumed to be structural ledges.* At the non-Burgers related interfaces, when the OR exhibits a relatively small deviation from the Burgers OR, both growth and structural ledges are again observed. Even when the deviation from the Burgers OR is large and ledge structures can no longer be observed with our Philips EM 420, planar facets are present, implying that a partially coherent interfacial structure exists at these interface as well. Curiously, no misfit dislocations were observed at either Burgers or non-Burgers α:β boundaries. During the present report period, emphasis was accordingly placed upon attempting to test the hypothesis that some of the ledges observed are of the structural type by a combination of TEM and modeling studies, and also upon understanding why misfit dislocations were not seen. Additionally efforts were made to broaden our "data base" of interfacial structure observations on grain boundary α allotriomorphs.

2.2. Experimental Procedures

The Ti-7.15 W/O Cr alloy was obtained from Titanium Metals Corporation of America, Henderson, NV. It was homogenized for 3 days at 1000°C. The homogenized alloy was then hot rolled by TIMET and the product was given the same homogenization treatment. It was immediately recognized that the coarse beta grain size characteristic of this alloy (and most other Ti-X alloys capable of being converted 100% to the beta phase) would make a TEM study of grain boundary alpha allotriomorphs infeasible. Dr. Donald Kroger of the Oak Ridge National Laboratory solved this problem for us by rapidly solidifying small portions of the homogenized alloy by means of the hammer and anvil technique. This yielded a beta grain size a few tens of microns in diameter and ensured that most thin areas, prepared for TEM, would contain portions of at least a few beta grain boundaries. Specimens cut from the rapidly quenched material were wrapped in Ta foil and encapsulated in Vycor under a vacuum ~ 10^{-5}

*Structural ledges compensate misfit by causing areas of good matching to repeat on their terraces and thereby markedly raise the average level of coherency at the interfaces (15). Although there is dispute about the mobility of these ledges, it seems unlikely at this time that their risers are able to migrate to a significant extent (16).

torr. They were transformed by upquenching directly to the intended isothermal reaction temperature. During this process, the omega phase precipitates must have dissolved prior to allotriomorph nucleation and growth, since it is a long familiar observation that in the presence of omega the alpha phase nucleates preferentially at omega:beta interfaces, yielding an ultra-fine dispersion of minute alpha crystals. TEM specimens were prepared by means of ion milling, since electropolishing has been convincingly shown to produce the "interface phase" (17), which grows allotriomorphically along alpha:beta boundaries and hence would entirely prevent this investigation from being conducted. The orientation relationships between alpha allotriomorphs and their bounding beta grains were determined by analyzing the Kikuchi pattern obtained from each phase. At least two different tilting conditions were used for a given phase. The error associated with this procedure was less than 1°.

2.3. Techniques Used to Model bcc:hcp Interfacial Structure

To model the structure of bcc:hcp boundaries, a computer-based graphical technique similar to that used by Rigsbee and Aaronson (18) for fcc:bcc boundaries is being employed. At the same time, Bollmann (19) O-lattice calculations of the misfit dislocation structure of these interfaces are being conducted. These two techniques are complimentary. Whereas the graphical technique, as presently employed, provides immediate visual impressions of atomic fit and misfit between two pre-selected rational planes, one from each lattice, as affected by variations in the ratio of their lattice parameters and the rotation of one plane with respect to the other, the Bollmann technique is applicable to any interface but requires appreciable calculation and even some judgment in order to secure "output".

Considering first the graphical modelling, a two-dimentional computer plot of all the atom positions is first prepared for parallel planes of a given lattice orientation relationship (OR). Fig. 1 shows examples of such plots for $(0001)_\alpha$ (Fig. 1 (a)) and $(0\bar{1}1)_\beta$ (Fig. 1(b)). For each combination, the two atom position plots can be superimposed with any desired rotation, though the structure of the interfaces is the same for all combinations, except for lateral shifts which do not change the basic geometry. One next seeks, again through computer programming, "coherent atom pairs", defined as atoms located on opposite sides of the α:β boundary which are displaced laterally with respect to one another by less than 15% of the average nearest neighbor distance in the two lattices. Structural ledges are now incorporated to increase the

fraction of the interfacial area made up from coherent atom pairs. These pairs are normally grouped in "patches", whose size and shape are sensitive to the precise details of lattice rotation and lattice parameter ratio. Construction of ledged interfaces begins with the selection of a "base" pair of coherent "patches". Monatomic structural ledges are then introduced parallel to the vector defined by the line connecting the best matching pair of atoms in each of these two patches. The spacing between structural ledges is adjusted to permit a contribution to interfacial coherency to be made by all combinations of interfacial planes. Interfaces with biatomic and higher ledges can be created by increasing the inter-ledge spacing proportionally and by allowing only the appropriate combinations of the interfacial planes to contribute to the interface. Knowing the direction, spacing and height of structural ledges, deviation of the apparent habit plane from the terrace plane can then be calculated. The line direction of misfit dislocations is finally established parallel to the vector between the best matching atom pairs in adjacent coherent patches on successive terraces of the ledged interface.

Independently of the graphical technique, Bollmann (19) O-lattice calculations were also made for the same interfaces to calculate their structure in terms of two arrays of misfit dislocations, i.e. without incorporating structural ledges. This method of calculation is summarized in the Appendix (section 2.7).

The lattice parameters used for both analysis techniques are as follows: a_β = 0.325 nm, a_α = 0.29564 nm and c_α = 0.46928 nm. They are appropriate to the α and β phases in the Ti-Cr alloy in the reaction temperature range used in this study. Table 1 shows the conjugate planes for the three best known ORs between bcc and hcp phases. Even though only the Burgers OR obtains in the Ti-Cr alloy being investigated, to give the modelling studies greater generality the interfacial structures associated with all three of these ORs are being investigated. Following Dahmen (20), the Potter (21) OR is obtained by a 1.38° rotation from the Burgers around the $[11\bar{2}0]_{hcp}$ axis; the Pitsch-Schrader (22) OR is given by a 5.26° rotation from the Burgers around the $[0001]_{hcp}$ axis.

2.4 Results

The 8 pairs of conjugate planes in Table 1 were taken to be conjugate habit planes and their structure is being examined, first by means of the graphical technique. Fig. 2(a) shows the

$(0001)_\alpha//(0\bar{1}1)_\beta$, $[11\bar{2}0]_\alpha//[111]_\beta$ interface of the Burgers OR. The coherent atom patch obtained is seen to be elongated along a particular direction which is approximately $[13\bar{4}0]_\alpha//[955]_\beta$. Two sets of misfit dislocations are required in this structure. Their spacings, 1.7 nm and 7.3 nm, are not only those between parallel misfit dislocations but also correspond to the spacings between adjacent parallel patches and between adjacent columns of patches, respectively. Fig. 2(b) shows the structure of the $(1\bar{1}00)_\alpha//(2\bar{1}1)_\beta$, $[11\bar{2}0]_\alpha//[111]_\beta$ interface for the Burgers OR. The coherent patches are nearly rectangular and two sets of misfit dislocations are present, forming a rectangular net whose spacings are 5.9 nm and 22.3 nm, respectively.

By a rotation of 5.26° around the axis normal to the same conjugate planes as were employed in constructing Fig. 2(a), the plot of Fig. 3(a), which shows the structure of the $(0001)_\alpha//(0\bar{1}1)_\beta$, $[11\bar{2}0]_\alpha//[111]_\beta$ interface for the Pitsch-Schrader OR, can be obtained. The coherent patches are now about equi-axed, and three sets of misfit dislocations, which compose a slightly elongated hexagonal net, are predicted. The spacing between parallel dislocations is 3.0 nm for one set, and 2.7 nm for each of the other two sets. Fig. 3(b) shows the structure of the $(01\bar{1}0)_\alpha//(011)_\beta$, $[0001]_\alpha//[0\bar{1}1]_\beta$ interface for the Pitsch-Schrader OR. Coherent patches are now elongated along the $[0001]_\alpha//[0\bar{1}1]_\beta$ direction. An even more elongated hexagonal network, formed by three sets of dislocations, is predicted for this interface, with spacings of 3.0, 7.0 and 7.0 nm, respectively.

For the Potter OR, the structure of the $(1\bar{1}01)_\alpha//(1\bar{1}0)_\beta$, $[11\bar{2}0]_\alpha//[111]_\beta$ interface is shown in Fig. 4. Large, elongated areas of coherent atom pairs are found. One of the two sets of dislocations required has a spacing of 2.8 nm; the other set has a much larger spacing, not shown in this Figure. The balance of the conjugate planes given in Table 1 was found to exhibit good matching only in a particular pair of directions but to provide a poor overall planar matching and, hence, such interfaces seem less likely to be the atomic habit plane for bcc:hcp interfaces.

To date, structural ledges have been incorporated only in the atomic conjugate planes for the Burgers OR. Fig. 5 shows a partially coherent interface with biatomic structural ledges whose terrace plane is $(1\bar{1}00)_\alpha//(2\bar{1}\bar{1})_\beta$. An excellent recurrence of coherent patches

associated with regularly spaced (ca. 1.0 nm) structural ledges can be seen. Comparison with Fig. 2(b) shows that introduction of biatomic structural ledges increases the proportion of the interfacial area which is coherent and eliminates one set of misfit dislocations whose Burgers vector is $a_\alpha/3[11\bar{2}0]$. This interface now needs only one set of misfit dislocations whose Burgers vector is $c_\alpha[0001]$ and spacing is 22.3 nm. Also, 4-atomic and 6-atomic height structural ledges were found to be successful in compensating misfit in the same direction as the bi-atomic ledges. Further, in the presence of 6-atomic height ledges, the interface can step both 'up' and 'down' while maintaining the same two sets of atom matching layers, whereas for bi- and 4-atomic height ledges, geometrical constraints limit the flexibility of the bcc:hcp interface by allowing stepping only in one direction (i.e. 'up' or 'down'). Structural ledge analyses for the other well known ORs are currently in progress.

O-lattice calculations have so far been made for two pairs of conjugate habit planes derived from the Burgers OR, $(0001)_\alpha//(0\bar{1}1)_\beta$, $[11\bar{2}0]_\alpha//[111]_\beta$ and $(1\bar{1}00)_\alpha//(2\bar{1}\bar{1})_\beta$, $[11\bar{2}0]_\alpha//[111]_\beta$. Open circles in Fig. 6 represent the O-points on the interface, i.e., the points of optimum matching between the two structures at the boundary. Although this analysis was performed to produce dislocation structures alternative to those determined by the graphical technique, as may be seen by comparing Fig. 6(a) with Fig. 2(a) and Fig. 6(b) with Fig. 2(b), an identical structure was obtained for each of these interfaces from the two techniques. Inasmuch as this result should be expected in the absence of structural ledges, it provides further support for the correctness of the graphical procedure as pointed out by Hall et al (23) for fcc:bcc interfaces. The Burgers vectors determined are $b_1 = a_\alpha/3[11\bar{2}0]$ and $b_2 = a_\alpha/3[\bar{1}2\bar{1}0]$ for the interface of Fig. 6(a), $b_1 = a_\alpha/3[11\bar{2}0]$ and $b_2 = c_\alpha[0001]$ for that of Fig. 6(b). Both sets of misfit dislocations in Fig. 6(a) are edge dislocations whereas those in Fig. 6(b) are both of mixed type. O-lattice calculations are also presently in progress for the other ORs.

2.5 Discussion

The TEM observations reported in the previous Interim Report appear to provide the first experimental evidence that ledge structures exist in the interfaces of GBAs formed at matrix grain faces with respect to both of the bounding matrix grains, and hence strongly imply that the interfaces of alpha GBAs are partially coherent whether or not there is a low energy OR with the

beta matrix grain, even though misfit dislocations have not been observed on them. These results contravene the long-held view on the interfacial structure of GBAs (8,9) but support that recently developed on the basis of the indirect evidence from nucleation and growth kinetics of allotriomorphs (7).

The absence of misfit dislocations on growth and structural ledge terraces raises the possibility that another important discovery may be materializing. The predicted inter-dislocation spacings are often an order of magnitude larger than the c. 2 nm. spacings observed on the broad faces of ferrite sideplates by means of the weak-beam, dark-field technique. Since this technique is also being employed during the present study, it seems unlikely that absence of misfit dislocations is only apparent. We are now beginning to consider the possibility that the misfit strain vector associated with structural ledges in the direction parallel to the terraces may be capable of replacing interfacial dislocations as a misfit-compensating mechanism. This possibility is considered below for a particular structural ledge configuration.

With respect to structural ledges, when Hall et al originally proposed the possibility of their presence, they recognized that these ledges can significantly increase the proportion of coherent areas in fcc:bcc interfaces. Rigsbee and Aaronson (24) found the predicted structural ledges (and misfit dislocations) on the broad faces of proeutectoid ferrite plates embedded in retained austenite in an Fe-C-Si alloy. However, structural ledges appear to be absent from faceted areas of fcc:bcc interfaces in a Cu-Cr alloy and also in a duplex stainless steel (though much less certainly), perhaps because these alloys were reacted for very long times prior to TEM observation. Recently, Hackney and Shiflet (25) have reported that structural ledges, as well as growth ledges and misfit dislocations, are present at facets on θ plates in an Al-Cu alloy. The current study based upon a graphical technique raises the possibility that structural ledges may also exist in bcc:hcp interfaces. Fig. 7 is an isometric sketch of a partially coherent bcc:hcp interface for the Burgers OR with structural ledges whose terrace is formed by $(\bar{1}100)_\alpha//(\bar{2}11)_\beta$, $[11\bar{2}0]_\alpha//[111]_\beta$.

Consider now the structure of the risers of structural ledges. In the presence of the Burgers OR, there appears to be two likely pairs of conjugate habit planes for these interface. One is $(01\bar{2}0)_\alpha//[111]_\beta$; this pair is perpendicular to the terrace plane. The second possibility is $(01\bar{1}0)_\alpha//[011]_\beta$. This interface forms an angle of 60°, rotated about $[001]_\alpha$, with the terrace.

Both of these interfaces can be modeled as a partially coherent boundary, at least for risers several atomic layers high. Hence it appears that unless kinks are formed on the risers of structural ledges so that they can be displaced by a ledge-on-ledge mechanism and the structural ledges can migrate synchronously, the risers should be immobile, as are those on fcc:bcc structural ledges (16).

One clue as to the cause for the absence of misfit dislocations on the ledged boundaries of α GBAs is provided by Fig. 8, showing a $(001)_\alpha$ projection of the bcc:hcp interface containing a bi-atomic structural ledge whose terrace is formed by $(1\bar{1}00)_\alpha//(2\bar{1}\bar{1})_\beta$ habit planes. The structural ledge is seen to have a Burgers vector of $a_\beta/6[111]$ (equivalent to $a_\alpha/9[11\bar{2}0]$). Hence this ledge can also perform a misfit compensating function parallel to the terrace planes. In the case of fcc:hcp interfaces, Shockley partial dislocations whose Burgers vector is parallel to the terrace plane serve as both misfit dislocations and as growth ledges for the $(111)_{fcc}//(0001)_{hcp}$, $[110]_{fcc}//[11\bar{2}0]_{hcp}$ interface (26). The image contrast exhibited by these dislocation-ledges is of the dislocation type. Dislocation-like defects showing image contrast corresponding to an a- or $[11\bar{2}0]_\alpha$-type Burgers vector have also been observed at bcc:hcp interfaces (27). Some of these defects, however, may actually be ledges, perhaps of the structural type, a few atomic layers high. High-resolution TEM will be required to resolve this question. An instrument of this type available at Case Western Reserve University will shortly be utilized for this purpose through the courtesy of Prof. Arthur Heuer.

More quantitative evidence supporting the presence of structural ledges in Burgers-related α:β interfaces can be derived from the apparent habit plane of "normal α" plates* in the same Ti-Cr alloy. Fig. 9 is a bright-field micrograph showing an edge-on image of the terrace of ledges, uniformly and closely spaced, on the broad faces of α plates formed at 700°C in a Ti-7.15 w/o Cr alloy. The zone axis operating is $[0001]_\alpha//[0\bar{1}1]_\beta$. The arrowheads in the Figure point out very small steps associated with ledges, whose height is 0.8 - 1.0 nm. From this micrograph and the corresponding diffraction pattern, the habit plane of those terraces was determined to be approximately $(\bar{5}7\bar{2}0)_\alpha//(\bar{9}88)_\beta$. This habit plane, obtained experimentally, is

*"Normal α" plates tend to be ill-formed, relatively thick and to lengthen with the assistance of much edge-to-edge sympathetic nucleation; "black-plates", formed at lower temperatures but also hcp with a Burgers OR with respect to β, are thin and well formed; the two types of plate are parallel to different habit planes (28,29).

within 2° of the apparent habit plane of the interface with structural ledges of which terrace is $(\bar{1}100)_\alpha//(\bar{2}11)_\beta$, as shown by the sketch in Fig. 7. The theoretical apparent habit plane is approximately $(\overline{11},\ 15,\ \bar{4},\ 0)_\alpha\ //(\overline{13},\ 11,\ 11)_\beta$.

Fig. 10 shows a $[11\bar{2}0]_{hcp}$ stereographic projection on which the directions of ledge risers determined by trace analysis are plotted (represented by + symbols), together with the apparent habit planes normals (denoted by filled circles) for the broad faces of the α plates determined stereographically as the cross product of the riser direction and the intersection of a broad face of the plate with the foil surface. It was again found to be near $(\overline{11},\ 15,\ \bar{4},\ 0)_\alpha//\ (\overline{13},\ 11,\ 11)_\beta$ (represented by an open square), with some scatter within the region enclosed with a dashed ellipse. Also, it was noted that the riser directions determined was near the line direction of c-type misfit dislocation in the interface with structural ledges shown in Fig. 7, i.e. $[11,\ 4,\ \overline{15},\ 0]_\alpha//[22,\ 13,\ 13]_\beta$ (denoted as an open triangle).

Subsequently, in order to determine the effective Burgers vector associated with these ledges, a g.b analysis was performed on same type of ledges, uniformly and closely spaced (ca. 11.0 nm apart) on the broad face of another α plate, shown in Fig. 11. The result, as is seen in Table 2, indicates that these ledges have a Burgers vector of $c[0001]_\alpha$, lying in the terrace plane of the ledges, i.e. $(\bar{1}100)_\alpha//(\bar{2}11)_\beta$. (The visibility of these features at $g = [\bar{2}200]_\alpha$ is probably due to the small component of misfit normal to the terrace plane.) Compared with the line direction determined in Fig. 10, they were found to be almost edge-oriented. The inter-ledge spacing measured (about 11.0 nm) corresponds to one half of the spacing of c-type misfit dislocations on the structurally ledged interface shown in Fig. 7. This is probably because the extra (0001) plane, which is located half-way up along the c-axis in the unit cell of the hcp lattice, contributes to actual misfit compensation.

Thse results obtained for the habit plane and the nature of defects at broad faces of α plates indirectly support the presence of structural ledges on a Burgers-related bcc:hcp interface, though such ledges have not yet been directly observed on these interfaces due to insufficient resolution. Also, these findings indicate that another set of ledges which is clearly observable, with an uniform spacing and lying in a well defined direction, is actually playing a role as misfit compensators on the broad faces of α plates. Similar analyses are currently being made on GBAs. However, the presence of multiple habit planes on allotriomorphs and small

deviations from the exact Burgers OR, which probably occurs to optimize coherency between a GBA and both of the abutting matrix grains, will make the problem appreciably more complicated.

2.6 Future Plans

First, and most important, TEM observation of the interfacial structure of α GBAs will be continued for various faceted interfaces. Especially, efforts will be made to determine the atomic habit planes of the terraces of ledges and to ascertain definitely whether or not structural ledges are present at Burgers (and at non-Burgers) related interfaces. If necessary, this effort will be supplemented with high-resolution TEM. Simultaneously, through quantitative g.b and trace analyses, the Burgers vectors and line directions of any defects observed will be determined. Secondly, qualitative observations will be made on the effects of reaction time at the highest and at the lowest reaction temperatures used for interfacial structure of α GBAs.

Graphical and O-lattice modeling of relevant bcc:hcp interfaces will be completed for the three best known ORs between bcc and hcp crystals. The possibility of structural ledges with the Pitsch-Schrader and the Potter ORs will be investigated. By changing lattice parameter ratios (a_{bcc}/a_{hcp} and c_{hcp}/a_{hcp}), the variation of interfacial structure with these factors will also be investigated. Then, the present calculations will be extended to non-Burgers related interfaces using the graphical technique and/or O-lattice theory. Finally, quantitative comparisons of the models with experimental results on α GBAs in a Ti-Cr alloy will be made insofar as the instruments and techniques available will allow.

With respect to the growth kinetics of α GBAs, Menon and Aaronson (5) found the thickening rate to be somewhat faster than D_v-control allows and thus suggested that these kinetics are accelerated by the simultaneous operation of the rejector plate (10) mechanism. However, Enomoto (30) recently showed, with his finite difference analysis of ledge-wise growth, that during the early stages of such growth, kinetics more rapid than those predicted for planar disordered boundaries are possible. Hence a re-evaluation of measured thickening kinetics will be attempted employing the average inter-growth ledge spacing, ledge height, and ledge geometry observed during the present study, in cooperation with Dr. Enomoto.

2.7 Appendix: O-lattice Construction for Burgers and Near-Burgers ORs

The hcp and the bcc lattices, designated by 1 and 2, respectively, can be related through a transformation matrix A. Following Bollmann (19), A is chosen such that, close to the origin, the nearest neighbor sites in the two lattices are related. The unit cells chosen in both lattice are illustrated in Fig. 12. This arrangement yields the following variant of the Burgers ORs between the hcp and the bcc lattices:

$$(0001)_\alpha // (0\bar{1}1)_\beta, \quad [11\bar{2}0]_\alpha // [111]_\beta.$$

In order to obtain maximum geometrical and dimensional similarity between the two lattices, the unit cell parameter listed in Table III were chosen.

The matrices $S^{(1)}$ and $S^{(2)}$, which have the X, Y and Z coordinates of the vectors of a, b and c (Table III) of the chosen hcp and bcc unit cells as their column vectors, can now be constructed. Then the transformation matrix A is given by:

$$A = RS^{(2)} (S^{(1)})^{-1}$$

where R is a rotation matrix which conveniently specifies the OR (for the Burgers OR, R is equal to I in the present study).

The (X,Y,Z) coordinates of the O-lattice are thus expressed as:

$$X^O = (1-A^{-1})^{-1} b^L = T^{-1} b^L$$

where b^L are three non-coplanar Burgers vectors of interfacial dislocations, so as to be equal to $S^{(1)}$.

3. MASSIVE TRANSFORMATION IN A Ag-26 A/O Al ALLOY

3.1 Introduction

This investigation will comprise the Ph.D. thesis of Mr. Yiwen Mou, who is a faculty member on leave from Chongqing University in the PRC. This study will presumably be the first in which our view (31) that the interphase boundaries of crystals produced during a massive transformation are frequently partially coherent, rather than disordered as originally predicted by Massalski (32), will be critically tested. The alloy composition we are utilizing for this investigation, Ag-26 A/O Al, has been previously studied in the present context by Plichta and Aaronson (2), who succeeded only in ascertaining orientation relationships of ξ_m crystals with respect to their matrix β grains, using the selected area electron channeling technique. Among

the 47 ξ_m crystals examined in this manner (all of which nucleated at β grain faces or edges), 46 were found to have a Burgers orientation relationships with respect to at least one of the β grains with which they were in contact. Irrational orientation relationships were found in many instances with respect to the other β grain(s), but the observation that facets appeared even on the non-Burgers interfaces strongly suggested that these interfaces were also likely to be partially coherent. Our attempts during that investigation to observe interphase boundary structure were frustrated, however, by an unfavorable texture of the foils (which caused too many interphase boundaries examined to lie nearly parallel to the electron beam, where their structure could not be observed) and, above all, by the precipitation, at room temperature, of the μ phase, evidently by means of another massive transformation. We plan to circumvent the latter problem in the present investigation by the use of a cold stage attachment available for our Philips 420 TEM, and the former by the use of different schedules of specimen preparation.

Since it is our belief that the massive transformation is as completely governed by the considerations of Gibbs and successors, particularly with regard to the role of crystallography in nucleation kinetics, as is precipitation from solid solution, we are anticipating that the interphase boundary structures observed on ξ_m crystals will be closely similar to those which Mr. Furuhara has been finding on the broad faces of grain boundary alpha allotriomorphs in Ti-7.15% Cr. Hence the TEM portions of this investigation should follow a path similar to that now being utilized for the Ti-Cr study.

3.2 Alloy Preparation

An ingot of Ag-24.4 A/O Al was prepared for us by the ALCOA Research Laboratories in ALCOA Center, PA. However, this alloy proved to be too brittle to roll, quite possibly because of rapid precipitation of the μ phase. The ingot was then remelted as a Ag-26 A/O Al alloy, comparable to the one previously employed. Little difficulty was now encountered in rolling out sheets of alloy 0.3 mm. thick. We would have preferred the 24.4 A/O Al composition because this one lies exactly at the congruent composition in this system and thus there could be no complaint about possible transformation with the fcc + hcp region; however, this proved impracticable--and we do not think that such complaints will be serious.

3.3 Heat Treatment of Specimens

In order to obtain samples suitable for TEM observation of interphase boundary structure produced by the massive transformation, we must have both massive crystals and retained β matrix in the samples, and also make the two phases have appropriate proportions and distribution. An ideal microstructure would be many very small massive crystals formed on grain boundaries with a retained β matrix. Due to the very fast kinetics of massive transformations, however, microstructures of this type are difficult to secure. Another problem arises from the very large β grain size due to the great proclivity for grain growth in the Ag-26 A/O Al alloy. This situation further complicates TEM sample preparation; in most samples we cannot find any grain boundaries within the thin areas.

We tried many different ways to heat treat specimens prepared by the ALCOA Research Laboratory. The samples are about 0.3 mm. thick. We found that quenching rate played an important role in determining the microstructure of this alloy. With faster quenching rates, the transformation favored is changed from equilibrium ξ precipitation to massive ξ_m and finally to martensite. The quenching rate depends upon both the quenching medium and the specimen thickness. With many combinations of these two factors we always obtained big massive crystals, sometimes connected with each other along the grain boundaries as networks; in some samples, no retained β′ was observed. When very thin specimens were quenched into iced brine, the size and distribution of massive crystals were somtimes improved, but martensite formed in many cases.

We have finally incorporated plastic deformation and recrystallization in our heat treatment cycle, and have thereby obtained much better microstructures and also reduced the β grain size. Specimens 0.3 mm. thick are heated in a salt bath to 700°C for 10 minutes, followed by iced brine quenching. Then they are rolled to 0.2 mm. thickness (33% reduction) and once again heated to 700°C or a little higher, but in some instances for only 5 seconds, and are finally quenched into room temperature water. With these treatments we obtained microstructures with very small massive crystals separately distributed along grain boundaries and a matrix of retained β phase. The grain size of the specimens is also reduced from larger than 1 mm. to about 0.1 mm. We also found that a slightly higher temperature than 700°C can lead to a smaller quantity of massive ξ_m and more retained β phase.

Although an even smaller grain size may be useful in TEM samples, massive transformations become difficult to control in specimens with too small a grain size. This was observed during heat treatment of thin ribbon specimens made by splat quenching directly from liquid state. These ribbons have a composition of Ag-24.4 A/O Al and a grain size less than 0.05 mm. Even with iced brine quenching, all of the high temperature β phase in almost all specimens transformed into massive ξ_m crystals without retained β phase. A possible explanation for this result is that a much high nucleation rate per unit specimen volume of ξ_m obtains due to the larger number densities of grain faces, and especially of grain edges and corners, in the small grained specimens, and a much shorter time is required for growing nuclei to impinge upon their neighbors. These two effects greatly accelerate overall transformation kinetics.

3.4 TEM Sample Preparation

In order to increase the probability that electron transparent areas in a sample will appear in microstructural regions containing small massive crystals on grain boundaries with retained β as the matrix phase, we first grind, polish and etch a specimen, and then observe its microstructure with a light microscope. The specimen is next lightly marked at the most favorable areas and then trimmed so that the best of these areas lie in the middle of the sample.

Two methods were used for TEM sample preparation; twin-jet electropolishing and ion milling. The following electrolyte and polishing conditions were found fairly suitable for twin-jet electropolishing: 5% perchloric acid, 10% glycerine and balance ethanol, held at -15°C, with the current range 10-15mA and voltage 20-25V. A problem arising with this method is that some massive crystals disappear from thin areas due to preferred electropolishing of interphase boundaries. Further modifications of this procedure may thus be required.

Ion milling is also being employed to make TEM samples. The initial sample thickness used is about 0.03 mm. Conditions utilized for milling are current - 0.3 mA, voltage - 5kV, and an incident angle of 15° for the first 2 hours and then 10° until perforation (requiring 3.5 more hours). Due to the different characteristics of electropolishing and ion milling, the concept of combining these two methods, i.e. first electropolishing almost to perforation and then ion milling, may be helpful for sample preparation.

3.5 Initial TEM observations

We are now at the point of beginning the TEM work. We have made some observations and are sure that there are two phases, presumably massive ξ_m and retained β, present in thin areas of the samples. There are also some grain boundaries in the thin areas. The latter fact indicates that the grain size of the samples is acceptable for TEM research.

3.6 Future Plans

Following determination of ORs between β and ξ_m, observations will be initiated on the structure of Burgers-related β:ξ_m interfaces in order to ascertain whether they have a predominantly ledged structure or instead one containing misfit dislocations. These initial observations will then be extended to non-Burgers interfaces. Quantitative studies will next be initiated of the interphase boundary structures found, first at Burgers and then at non-Burgers-related interfaces, following overall a condensed version of the research pattern developed during the Ti-Cr investigation recounted in the previous section of this report.

4. SHEAR VS. DIFFUSIONAL MECHANISMS OF PRECIPITATE GROWTH

On present tentative plans, once the experimental investigations by graduate students of the interphase boundary structure of grain boundary alpha allotriomorphs in a Ti-Cr alloy and of massive:matrix interfaces in a Ag-Al alloy have been completed, it is planned to change somewhat the approach of this program. The question of shear vs. diffusional growth mechanisms has been either an underlying or a clearly visible theme of most of the individual research programs conducted with the support of this grant. We are now considering an effort, to be conducted primarily by the P.I. rather than by graduate students, to discern the fundamental differences between these two mechanisms of growth and to understand why, again in fundamental terms, such severe conflict has developed between their proponents. Interphase boundary structure will almost certainly be central to this effort as well.

Although this grant has for some time supported, "as a sideline", preparation of a number of invited papers dealing more or less directly with this issue, the P.I.'s personal effort has recently shifted more intensively in this direction. He is presently leading prepartion of a lengthy paper for the proceedings (to be published in Met. Trans. A) of the Interface Science and

Engineering Symposium held at the World Materials Congress in Chicago in which the wide range of individual issues comprising this controversy is being tackled "head on", in at least preliminary fashion. This effort will help considerably in consolidating the agenda for further work by the P.I. on this subject. In addition, preparation of the introductory paper for the International Conference on Bainite, held during the same Meeting in Chicago, also involves dealing with some of these issues as well as with a number of others less closely related.

CONSULTING ACTIVITIES WITH AFWAL, WRIGHT-PATTERSON AIR FORCE BASE

Although the schedule of this continuing effort has been somewhat disrupted by continuing problems with organization of reimbursement for the P.I.'s travel expenses to and from Wright-Patterson, the P.I. continues his interactions with members of the Structural Metals Branch of AFWAL. He has recently attempted to develop an arrangement with AFWAL, through Dr. Ted Nicholas, acting Titanium Technical Area Manager, for one of his former students, Prof. William T. Reynolds, Jr. (who is now with Virginia Polytechnic Institute, Blacksburg, VA), to tackle via computer preliminary screening of fiber compositions for the fiber-composite strengthening of titanium aluminides. The P.I. has been advocating just this approach to his colleagues at Wright-Patterson for several years. There are two technical ideas underlying this strategy. One is that because of the high solubility of Ti for so many elements it would be appropriate to ascertain via phase diagram calculations (rather than by means of costly and time-consuming experiments) whether or not a stable two- or multi-phase field exists between these aluminides and the proposed matrix (aluminide-based) composition. The P.I. feels that the concept of "diffusion barriers" is largely illusory; for high operating temperature applications, it seems better to trust thermodynamics rather than to kinetics. The second is that a useful preliminary assessment of the coarsening kinetics of fibers can also be made by computer using elementary coarsening theory. Fiber compositions which fail this test also should not be candidates for experiments.

It was thus encouraging to find that AFWAL is presently supporting an effort of this type with a small firm in the San Diego area; perhaps the P.I.'s past urgings have had some effect after all. The suggestion was offered that arrangements might be modified to permit the participation of Prof. Reynolds; no indications have as yet been received, however, as to whether or not this idea has been followed up.

PUBLICATIONS, HONORS AND AWARDS

AFOSR-supported Papers Listed as Being "In press" in the Previous Interim Report

H. I. Aaronson and W. T. Reynolds,Jr., "Reply to a Discussion by J. W. Christian and D. V. Edmonds of Papers by Aaronson and Co-Workers on the Proeutectoid Ferrite and Bainite Reactions", Scripta Met., 22, 567 (1988).

H. J. Lee and H. I. Aaronson, "Eutectoid Decomposition Mechanisms in Hypoeutectoid Ti-X Alloys", Jnl. Materials Science, 23, 150 (1988).

H. I. Aaronson and T. Furuhara, "Fundamentals of First Order Diffusional Phase Transformations: A Physical Metallurgical Viewpoint:, 'Analytical Electron Microscopy-1987', San Francisco Press, San Francisco, CA, p. 3 (1987).

H. I. Aaronson and R. V. Ramanujan, "Some Problems for Physicists in First Order Diffusional Phase Transformations in Crystalline Solids", on "Competing Interactions and Microstructures: Statics and Dynamics", Springer-Verlag, Berlin, FRG, p. 30 (1988).

M. Enomoto, W. T. Reynolds, Jr. and H. I. Aaronson, "Influence of Alloying Element Distribution Within and Very Near α:γ Boundaries upon Ferrite Growth Kinetics and Composition in Fe-C-X Alloys", 'Analytical Electron Microscopy-1987', San Francisco Press, San Francisco, CA, p. 65 (1987).

H. I. Aaronson and W. T. Reynolds, Jr., "The Bainite Reaction", 'Phase Transformations '87', Inst. of Metals, London,

W. T. Reynolds, Jr., H. I. Aaronson, F. Z. Li and C. K. Shui, "Formation of Granular Bainite in Fe-C-Mo Alloys", 'Phase Transformations "87', Inst. of Metals, London,

H. I. Aaronson and W. T. Reynolds, Jr., "Rejoinder to Comments by Christian and Edmonds", Scripta Met., 22, 575 (1988).

H. J. Lee and H. I. Aaronson, "Surface Relief Effects Associated with Proeutectoid Alpha Plates in Ti-7.15 W/O Cr Alloy", Acta Met., 36, 787 (1988).

H. J. Lee, G. Spanos, G. J. Shiflet and H. I. Aaronson, "Mechanisms of the Bainite (Non-

Lamellar Eutectoid) Reaction and a Fundamental Distinction between the Bainite and Pearlite (Lamellar Eutectoid) Reactions", Acta Met., 36, 1129 (1988).

H. J. Lee and H. I. Aaronson, "Morphology, Crystallography, Growth Kinetics and Mechanism of Bainite Nodules in a Hypereutectoid Ti-25 W/O Cr Alloy", Acta Met., 36, 1141 (1988).

H. J. Lee and H. I. Aaronson, "Re-examination of a Critical Experiment on Conditions for Formation of Bainite and Pearlite", Acta Met., 36, 1155 (1988).

6.2 AFOSR-supported Papers Since Published or Accepted for Publication

H. I. Aaronson, J. M. Rigsbee, T. Furuhara, N. Prabhu, W. T. Reynolds, Jr. and J. M. Howe, "Further Rebuttal to J. W. Christian and D. V. Edmonds", Scripta Met., in press.

T. Furuhara, A. M. Dalley and H. I. Aaronson, "Interfacial Structure of Grain Boundary Allotriomorphs in a Hypoeutectoid Ti-Cr Alloy", Scripta Met., 22, 1509 (1988).

S. K. Liu, W. T. Reynolds, Jr., H. Hu and H. I. Aaronson, "Comments on Reply to Discussion of 'The Bainite Transformation in a Silicon Steel'", Met. Trans., in press.

T. Furuhara and H. I. Aaronson, "On the Mechanisms of Interphase Boundary Carbide Precipitation", Scripta Met., 22, 1635 (1988).

H. I. Aaronson, M. Enomoto, T. Furuhara and W. T. Reynolds, Jr., "Influence of the Structure and Chemistry of α:γ Boundaries upon Grain Boundary Allotriomorph Growth Kinetics and Composition in Fe-C and Fe-C-X Alloys", International Conference on Physical Metallurgy of Thermomechanical Processing of Steels and Other Metals, Proceedings Vol. 1, p. 80, The Iron and Steel Institute of Japan, Tokyo (1988).

H. I. Aaronson, W. T. Reynolds, Jr. and G. Spanos, "Discussion of 'Low Temperature Ageing of Fe N Austenite' by J. Foct, P. Rochegude and A. Hendry. I. Mechanism of the Bainite Reaction", Scripta Met., in press.

M. Enomoto and H. I. Aaronson, "Influence of the Ledge Mechanism on Diffusivities Back-Calculated from the Migration Kinetics of Planar Interphase Boundaries in Two-Phase Diffusion Couples", Scripta Met., in press.

Yiwen Mou and H. I. Aaronson, "The Carbon-Carbon Interaction Energy in Alpha Fe-C Alloys", Acta Met., in press.

Raju V. Ramanujan, H. I. Aaronson and P. H. Leo, "The Role of Boundary Conditions in Modelling the Elastic Fields Around a Misfitting Precipitate", Met. Trans., in press.

H. I. Aaronson, D. Eylon, C. M. Cooke, M. Enomoto and F. H. Froes, "The Widmanstatten-Start (W_s) Temperature, as Affected by Matrix Grain Size, in Ti-6% Al-4% V and in Fe-C Alloys", Scripta Met., in press.

6.3 Honors and Awards

The P.I. has been invited by Mr. Alexander Scott, Executive Director of TMS, to serve as the 1990 Institute of Metals Lecturer and to receive the R. F. Mehl Award at the 119th Annual Meeting of TMS in Anaheim, CA during the period February 19-22, 1990.

REFERENCES

1. M. R. Plichta, J. C. Williams and H. I. Aaronson: Met. Trans., 8A, 1885 (1977).
2. M. R. Plichta and H. I. Aaronson: Acta Met., 28, 1041 (1980).
3. W. F. Lange III, M. Enomoto and H. I. Aaronson: Met. Trans., 19A, 427 (1988).
4. M. Enomoto and H. I. Aaronson: Met. Trans., 17A, 1385 (1986).
5. E. S. K. Menon and H. I. Aaronson: Met. Trans., 17A, 1703 (1986).
6. H. I. Aaronson and K. C. Russell: Proc. of an International Conference on Solid-Solid Phase Transformations, p. 371, TMS-AIME, Warrendale, PA (1983).
7. J. H. van der Merwe: Jnl. App. Phys., 34, 177, 123 (1963).
8. C. S. Smith: Trans. ASM< 45, 533 (1953).
9. H. I. Aaronson: Decomposition of Austenite by Diffusional Processes, p. 387 Interscience, NY (1962).
10. H. B. Aaron and H. I. Aaronson: Acta Met., 16, 789 (1968).
11. J. R. Bradley, J. M. Rigsbee and H. I. Aaronson: Met. Trans., 8A, 323, (1977).
12. A. D. King and T. Bell: Met. Trans., 6A, 1428 (1975).
13. W. G. Burgers: Physics, 1, 561 (1934).
14. T. Furuhara, A. M. Dalley and H. I. Aaronson: Scripta Met., 22, 1509 (1988).
15. M. G. Hall, H. I. Aaronson and K. R. Kinsman: Surface Science, 31, 257 (1972).
16. J. M. Rigsbee, E. S. K. Menon, H. J. Lee and H. I. Aaronson: Scripta Met., 17, 1465 (1983).
17. D. Banerjee, C. G. Shelton, B. Ralph and J. C. Williams: Acta Met., 36, 125 (1988).

18. J. M. Rigsbee and H. I. Aaronson: Acta Met., 27, 351 (1979).

19. W. Bollmann: Crystal Defects and Crystalline Interfaces, Springer Verlag, Berlin (1970).

20. U. Dahmen: Acta Met., 30, 63 (1982).

21. D. I. Potter: J. Less Common Metals, 31, 299 (1973).

22. W. Pitsch and A. Schrader: Arch. Eisenhuttenwesen, 29, 715 (1958).

23. M. G. Hall, J. M. Rigsbee and H. I. Aaronson: Acta Met., 34, 1419 (1986).

24. J. M. Rigsbee and H. I. Aaronson: Acta Met., 27, 365 (1979).

25. S. J. Hackney and G. J. Shiflet: Acta Met., in press.

26. C. Laird and H. I. Aaronson: Acta Met., 15, 73 (1967).

27. E. S. K. Menon and H. I. Aaronson: Acta Met., 34, 1975 (1986).

28. H. I. Aaronson, W. B. Triplett and G. M. Andes: Trans. TMS-AIME, 209, 1227 (1957).

29. E. S. K. Menon and H. I. Aaronson: Acta Met., 34, 1963 (1986).

30. M. Enomoto: Acta Met., 35, 935, 945 (1987).

31. H. I. Aaronson, C. Laird and K. R. Kinsman: Scripta Met., 2, 259 (1968).

32. T. B. Massalski: Phase Transformations, p. 433, ASM, Metals Park, OH (1970).

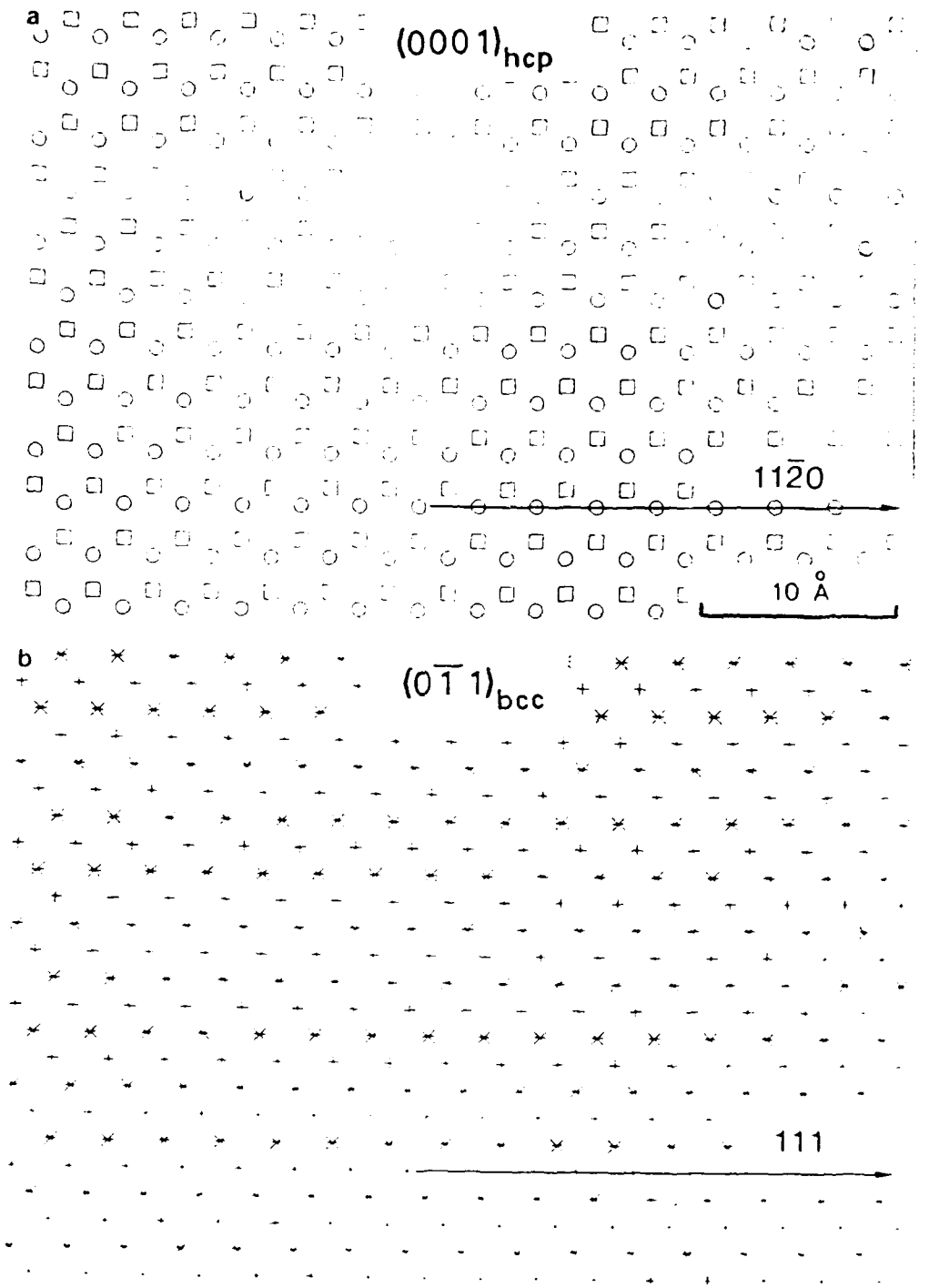


Figure 1: (a) Computer plot of hcp atom positions projected onto a $(0001)_{hcp}$ plane. o -- atoms on A layer. □ -- atoms on B layer. (b) Computer plot of hcp atom positions projected onto a $(0\bar{1}1)_{bcc}$ plane. + -- atoms on C layer. ✳ -- atoms on D layer.

Table 1: Well-known ORs between the bcc and hcp phases and the conjugate planes

Orientation relationship	Conjugate planes hcp // bcc
Burgers	(0001) // $(0\bar{1}1)$
	$(1\bar{1}00)$ // $(2\bar{1}\bar{1})$
	$(11\bar{2}0)$ // (111)
Pitsch-Schrader	(0001) // $(0\bar{1}1)$
	$(0\bar{1}10)$ // (011)
	$(2\bar{1}\bar{1}0)$ // (100)
Potter	$(1\bar{1}01)$ // $(1\bar{1}0)$
	$(11\bar{2}0)$ // (111)

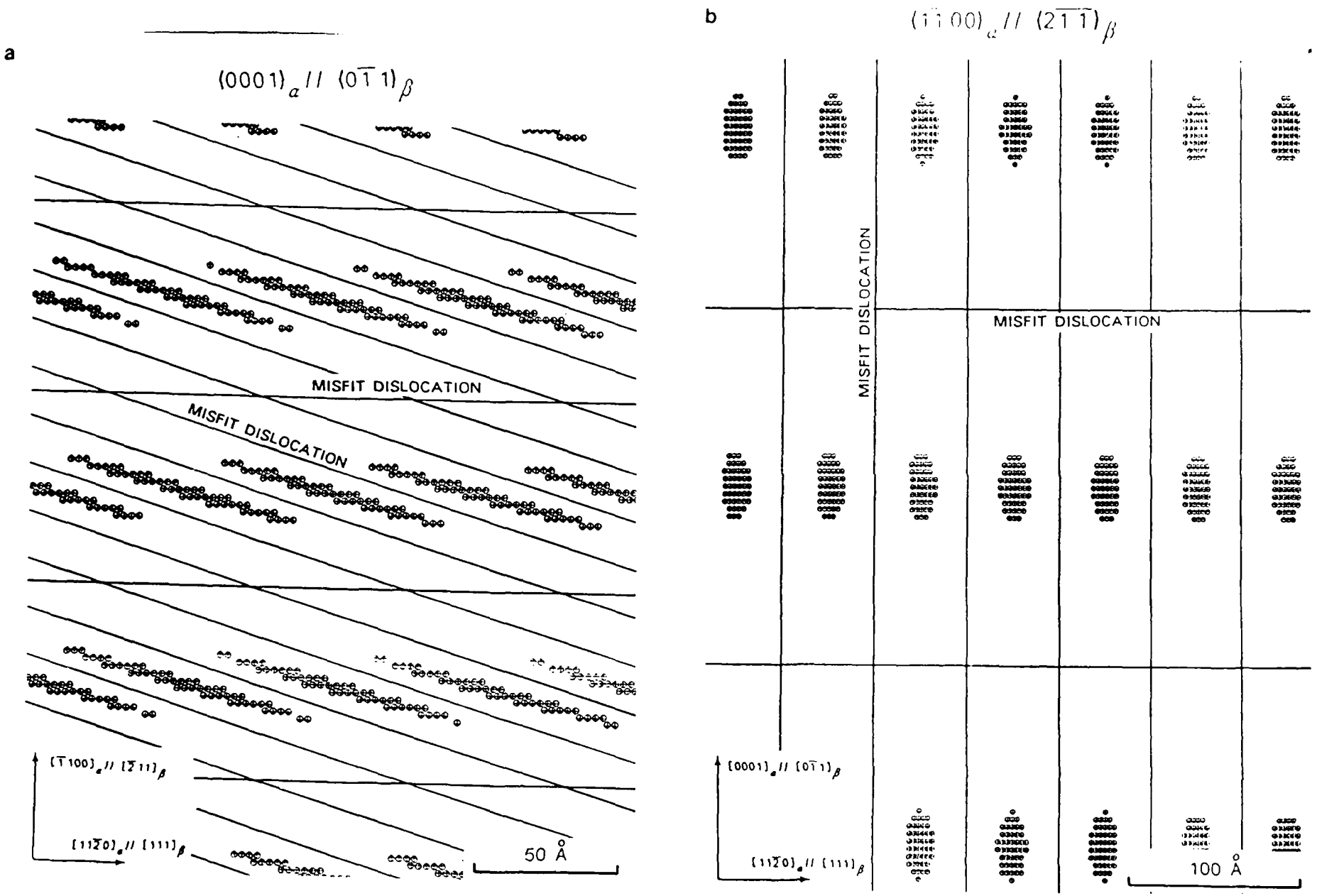


Figure 2: Computer plot of coherent patches and misfit dislocation structures on the Burgers-related interfaces. (a) $(0001)_\alpha$// $(0\bar{1}1)_\beta$, $[11\bar{2}0]_\alpha$// $[111]_\beta$ interface. (b) $(1\bar{1}00)_\alpha$// $(2\bar{1}\bar{1})_\beta$, $[11\bar{2}0]_\alpha$// $[111]_\beta$ interface. o -- hcp atoms. + -- bcc atoms.

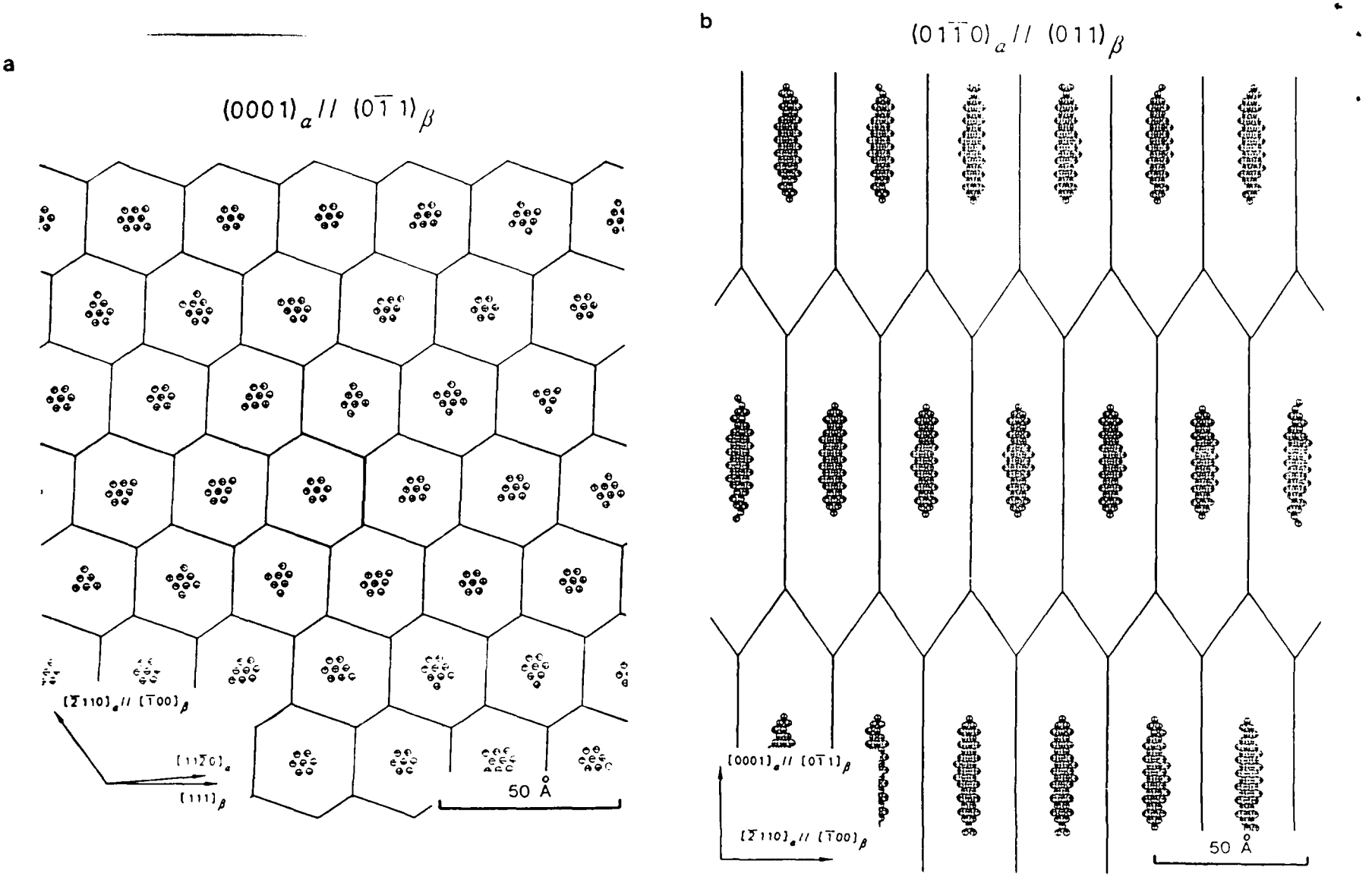


Figure 3: Computer plot of coherent patches and misfit dislocation structures on the Pitsch-Schrader related interfaces. (a) $(0001)_\alpha$// $(0\bar{1}1)_\beta$, $[2\bar{1}\bar{1}0]_\alpha$// $[100]_\beta$ interface. (b) $(01\bar{1}0)_\alpha$// $(011)_\beta$, $[2\bar{1}\bar{1}0]_\alpha$// $[100]_\beta$ interface. o -- hcp atoms. + -- bcc atoms.

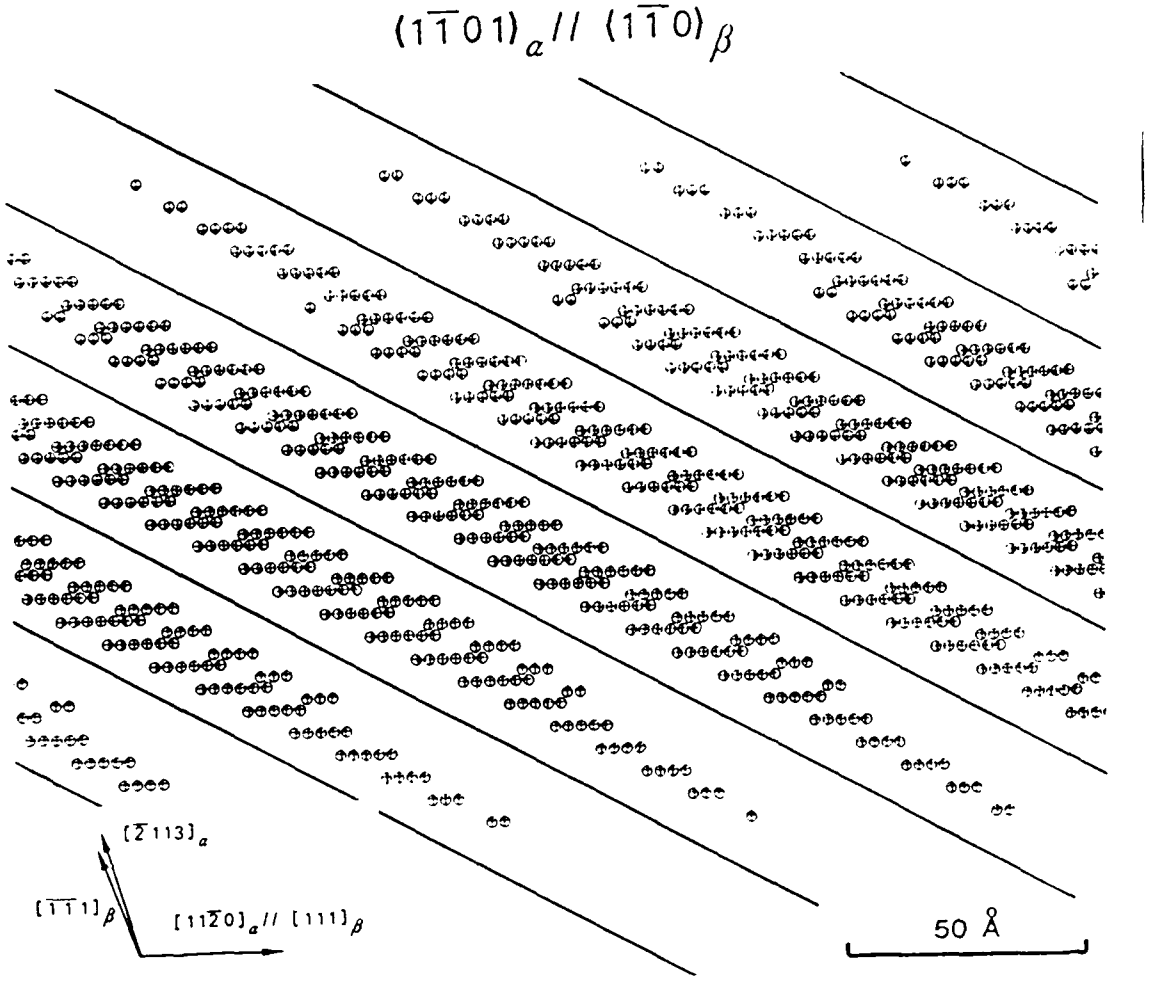


Figure 4: Computer plot of coherent patches and misfit dislocation structures on the $(1\bar{1}01)_\alpha$// $(1\bar{1}0)_\beta$, $[11\bar{2}0]_\alpha$// $[111]_\beta$ interface of the Potter OR. o -- hcp atoms. + -- bcc atoms.

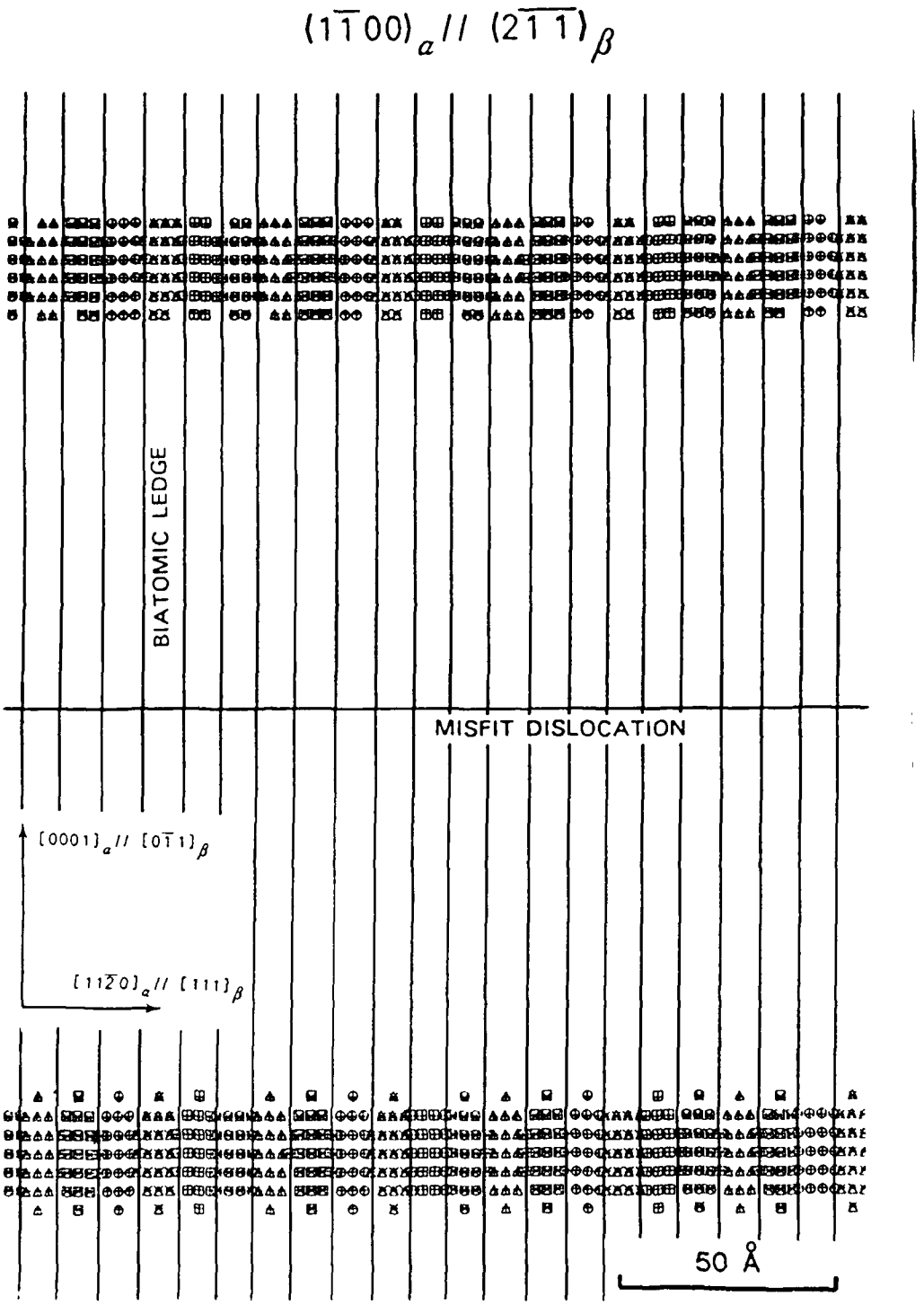


Figure 5: Computer plot of coherent patches and misfit dislocation structures with biatomic structural ledges of which the terrace is the $(1\bar{1}00)_\alpha$// $(2\bar{1}\bar{1})_\beta$, $[11\bar{2}0]_\alpha$// $[111]_\beta$ interface of the Burgers OR. o,□ , △ -- hcp atoms. +,✳-- bcc atoms.

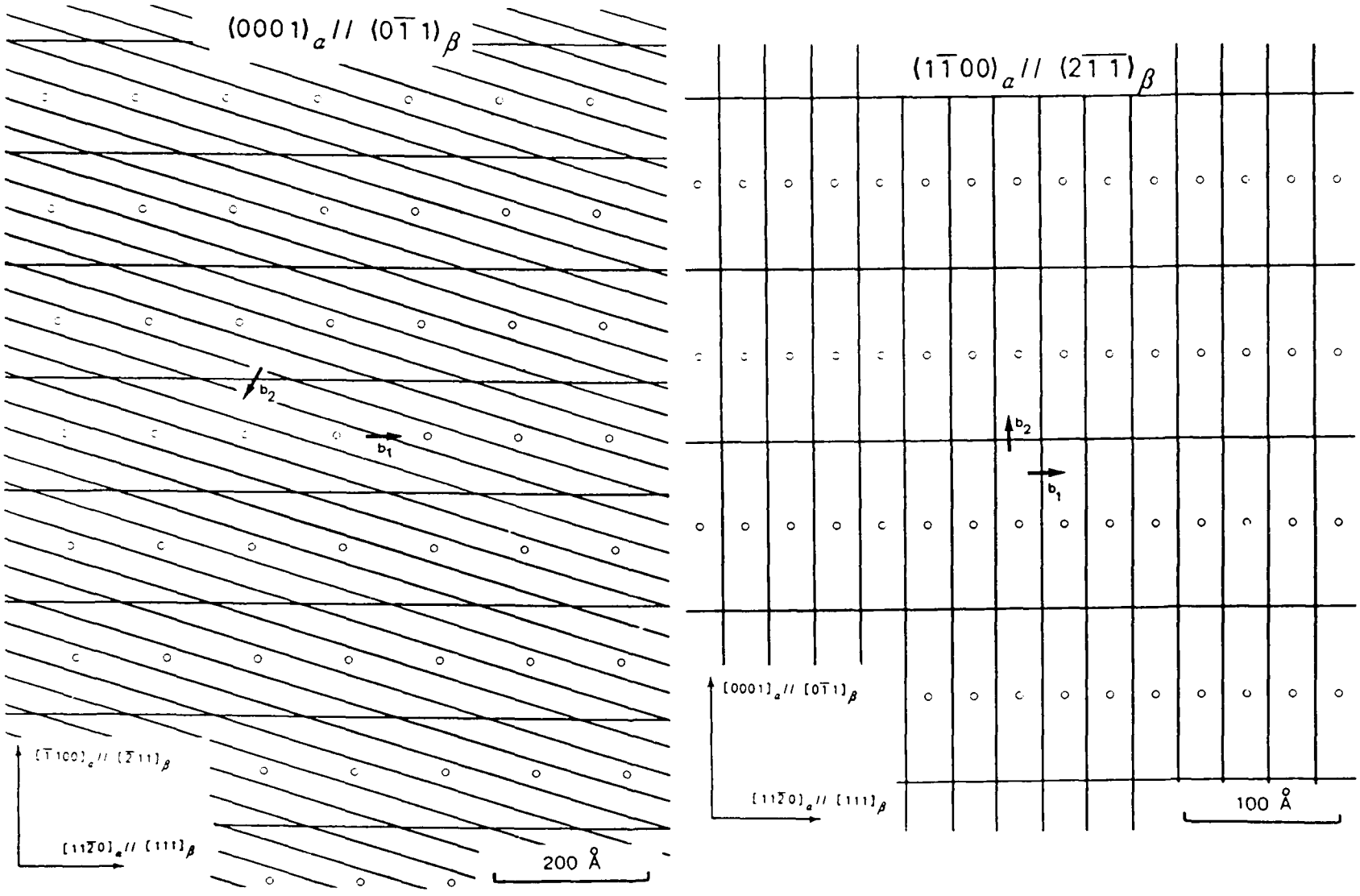


Figure 6: Computer plots of O-points and misfit dislocation structures on the conjugate planes of the Burgers OR. $\mathbf{b}_1$ and $\mathbf{b}_2$ are the Burgers vectors for the two sets of parallel dislocations, respectively. (a) the $(0001)_\alpha // (0\bar{1}1)_\beta$, $[11\bar{2}0]_\alpha // [111]_\beta$ interface. (b) the $(1\bar{1}00)_\alpha // (2\bar{1}\bar{1})_\beta$, $[11\bar{2}0]_\alpha // [111]_\beta$ interface.

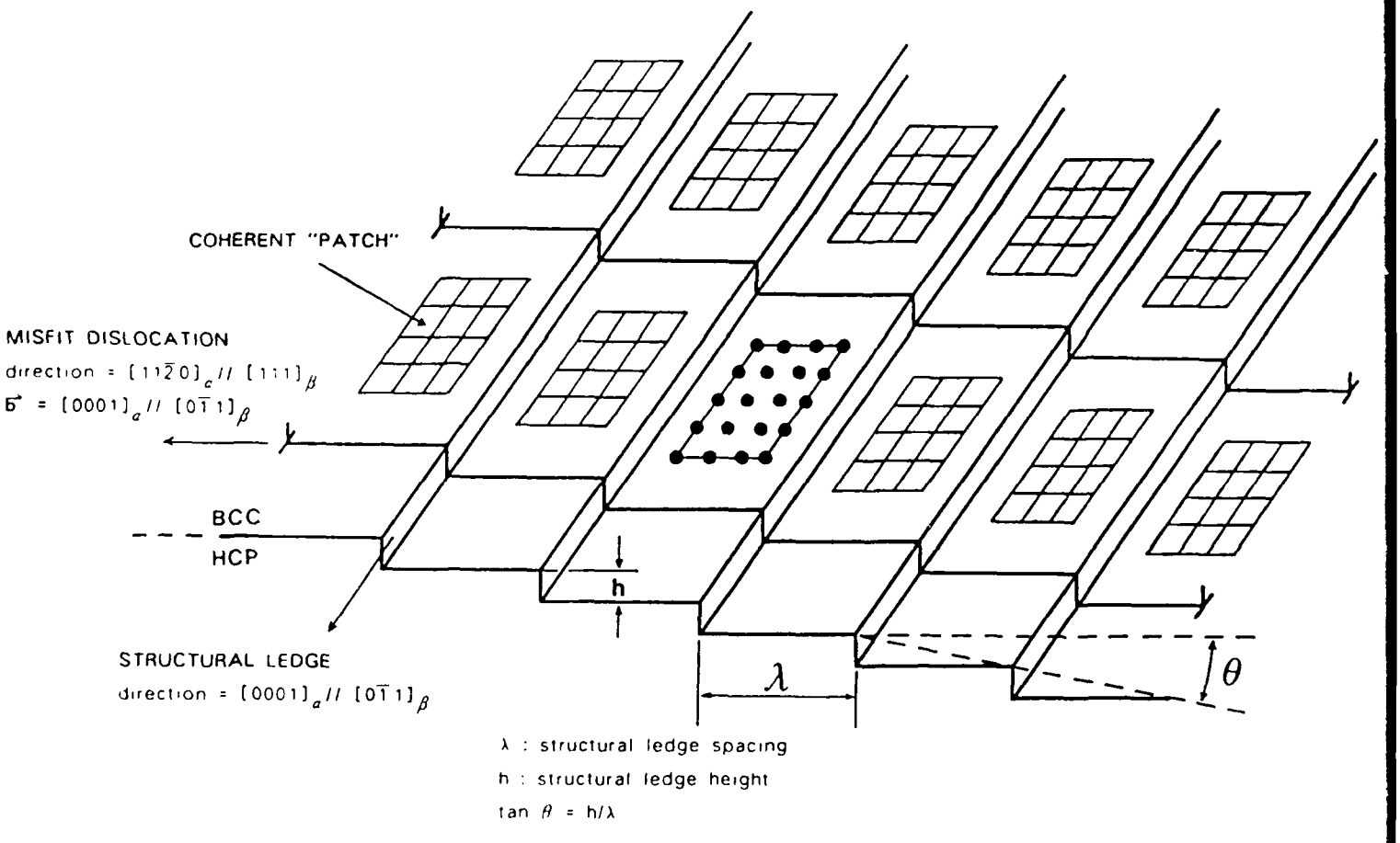


Figure 7: Isometric sketch of a Burgers-related bcc:hcp interface with structural ledges, of which the terrace plane is $(\bar{1}100)_\alpha$// $(\bar{2}11)_\beta$, and with misfit dislocations.

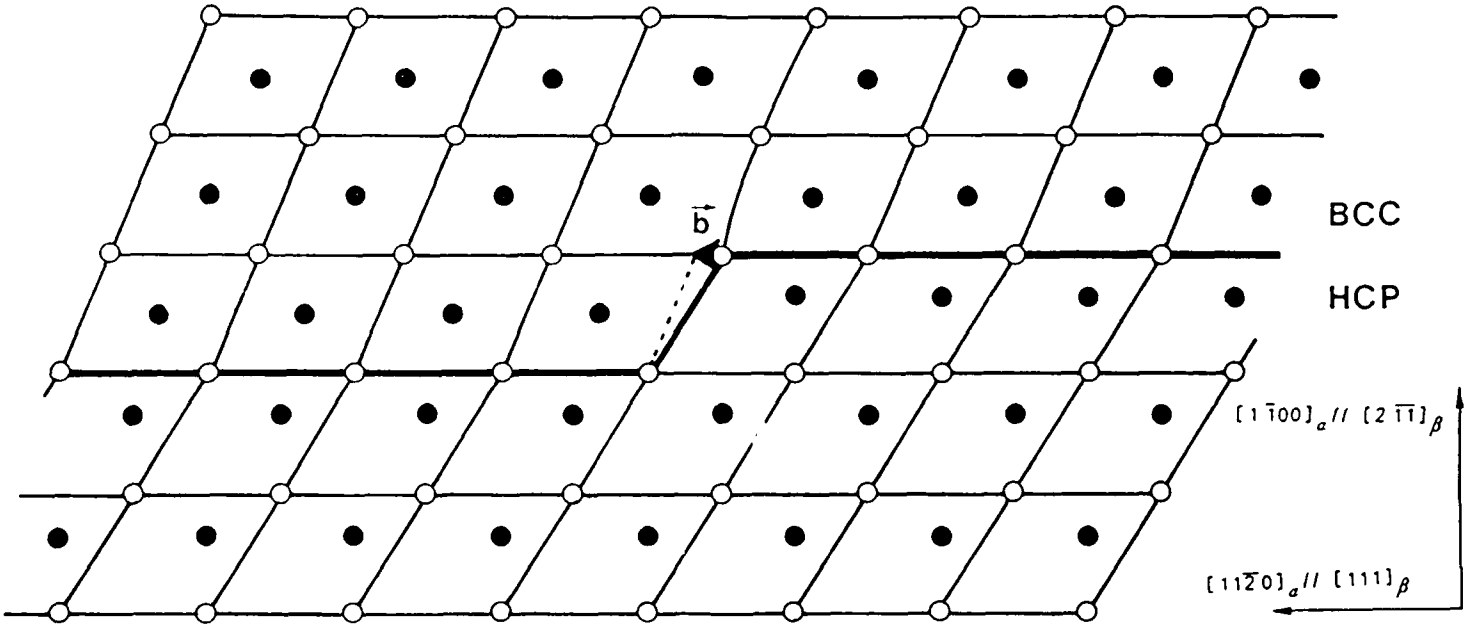
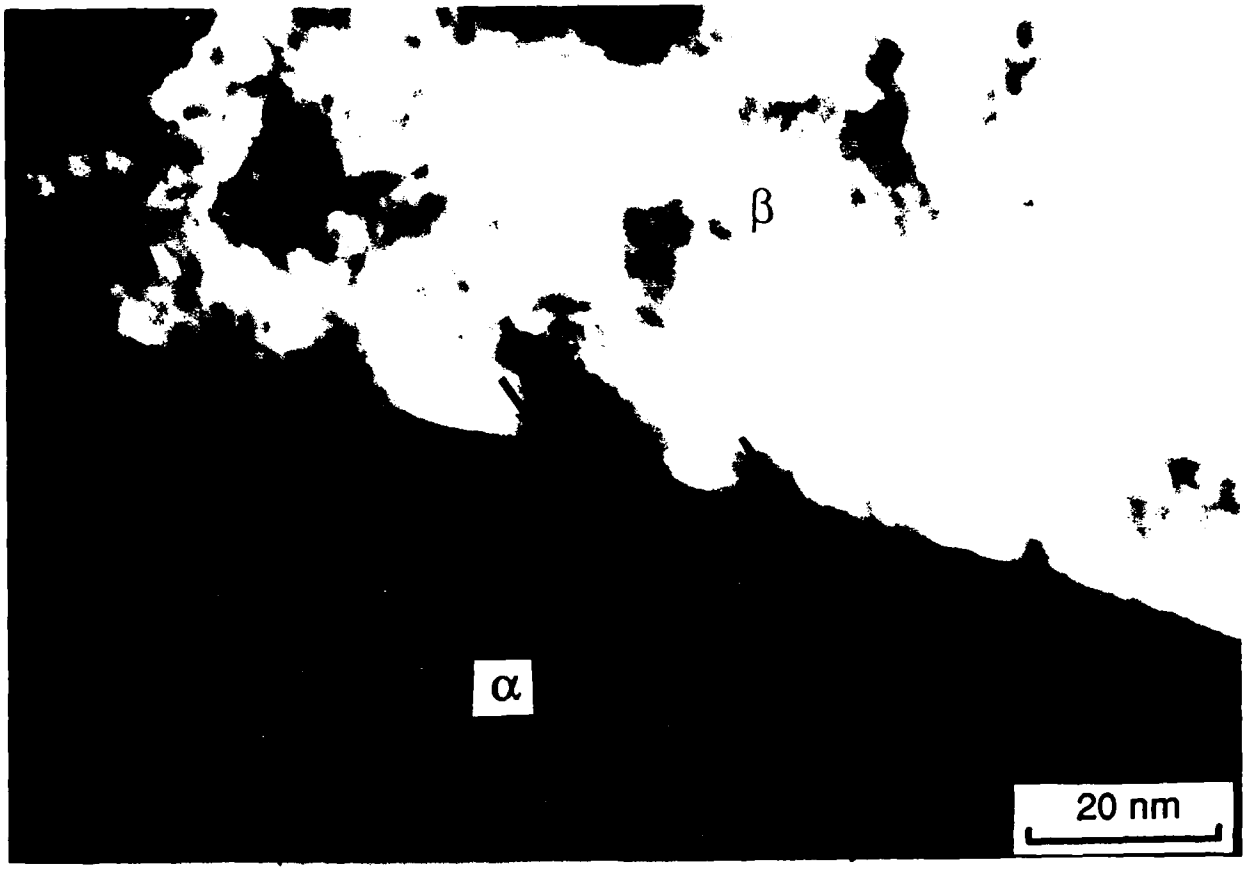
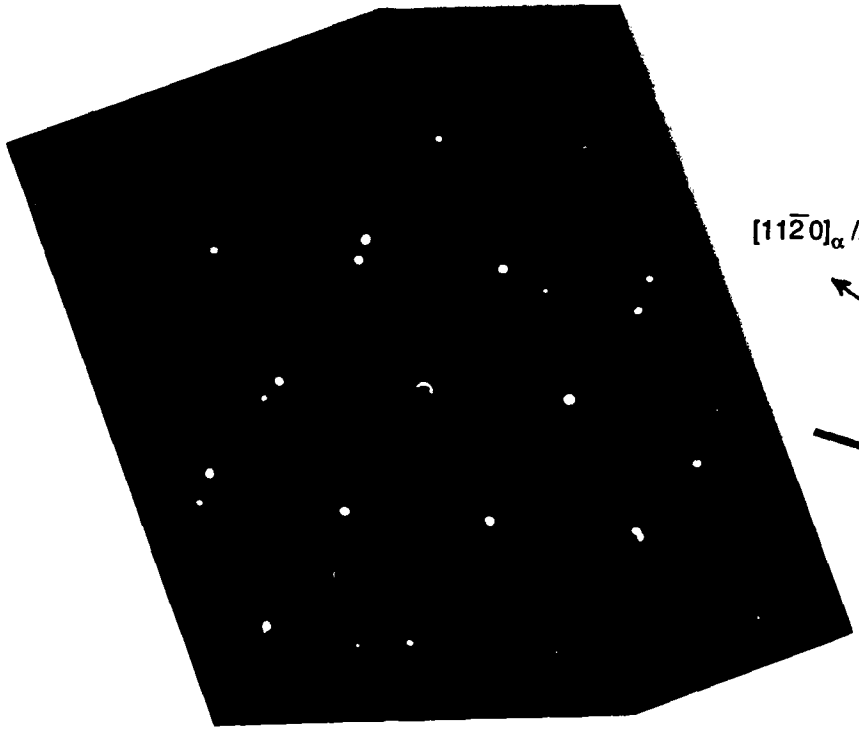
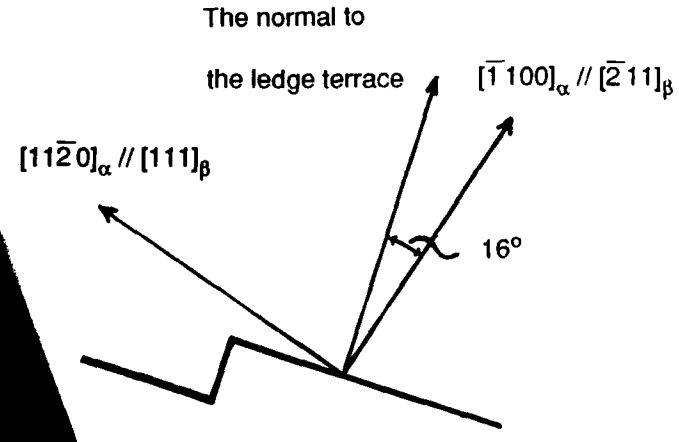


Figure 8: Schematic of a Burgers-related bcc:hcp interface projected onto an $(0001)_\alpha // (0\bar{1}1)_\beta$ plane, containing one biatomic structural ledge of which the terrace is parallel to the $(1\bar{1}00)_\alpha // (2\bar{1}\bar{1})_\beta$ plane, showing that the structural ledge is associated with a Burgers vector of $1/6\,[111]_{bcc}$.







$Z = [0001]_\alpha$ // $[0\bar{1}1]_\beta$

Figure 9: Bright-field micrograph showing small steps on the broad face of an intragranular α plate formed from β matrix in a Ti-7.15 w/o Cr alloy reacted for 1200 s at 873 K. The corresponding SAD pattern shows that this plate has a Burgers OR with respect to the matrix grain, and the habit plane of ledge terrace observed is approximately 16° away from $(\bar{1}100)_\alpha$// $(\bar{2}11)_\beta$.

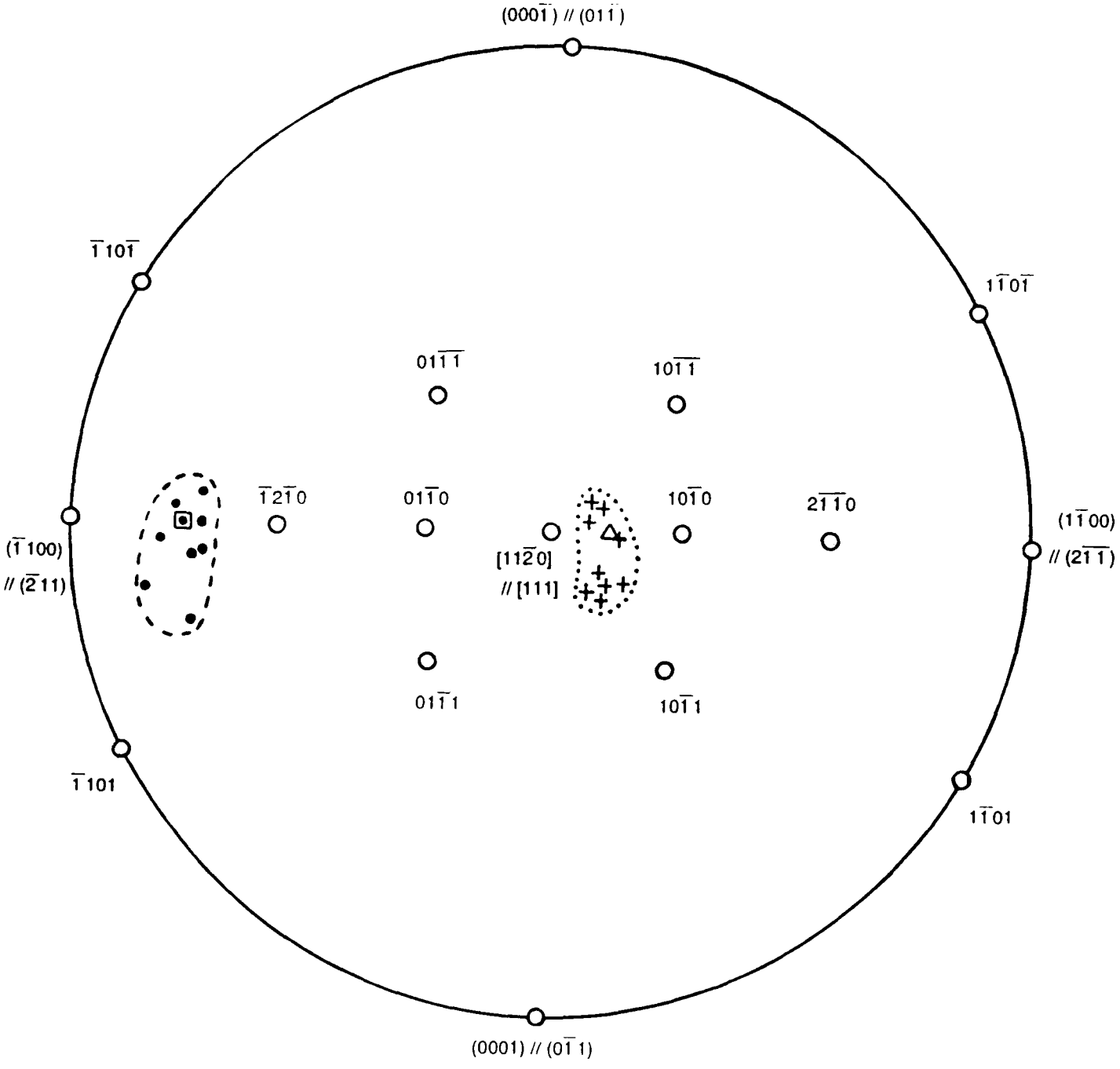


Figure 10: $(11\bar{2}0)_{hcp}$ stereographic projection on which are plotted the ledge directions on the broad faces of α plates, determined by trace analysis (+) and the observed habit planes (•) of the broad faces. The scatter of the ledge directions and the apparent habit planes of broad faces determined by trace analysis are shown by the ellipses defined by the dotted and dashed curve, respectively. Also, the habit plane of the interface with structural ledges (□) and the direction of c-type misfit dislocation (Δ) predicted for the interface in Fig. 7 are plotted.

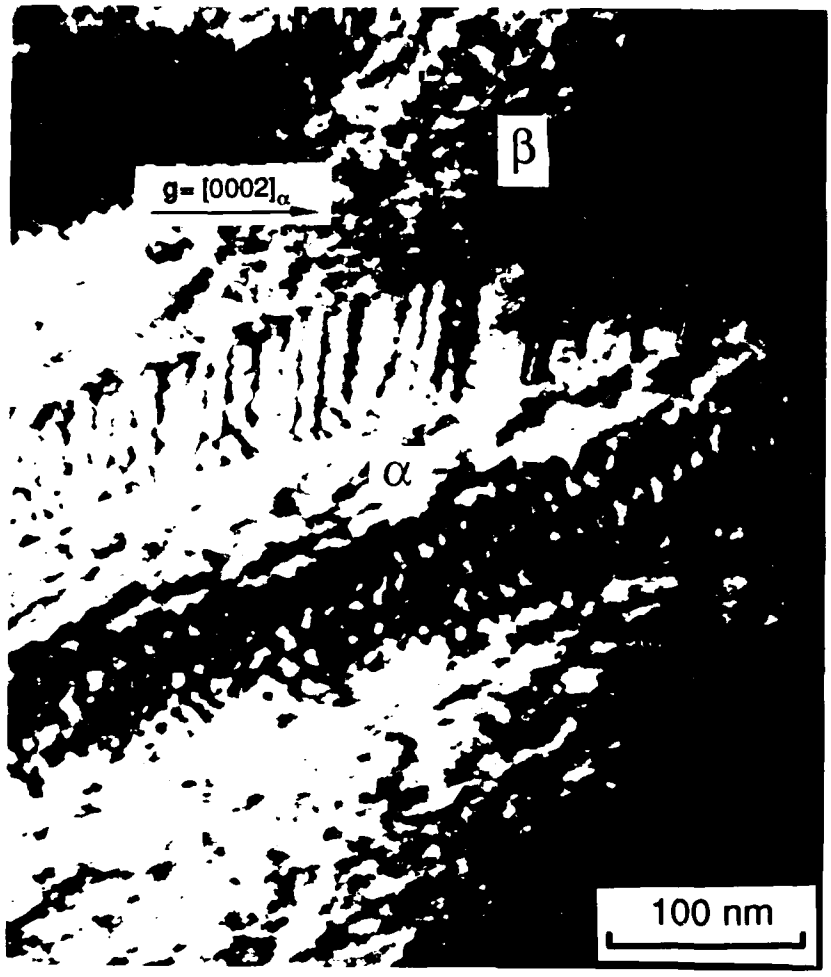


Figure 11: Dark-field micrograph showing fine ledges, uniformly spaced, on the broad face of another intragranular α plate. $\mathbf{g} = [0002]_{\alpha}$.

Table 2: Results of **g.b** analysis performed on the same ledges in Fig. 11.

$\mathbf{g}_{hcp}$	Observed Contrast	**g.b** for $\mathbf{c}_{\alpha}$[0001]
0002	strong	2
$01\bar{1}\,\bar{1}$	weak	-1
$02\bar{2}\,0$	none	0
$01\bar{1}\,1$	weak	1
$1\bar{2}10$	none	0
$\bar{1}\,101$	weak	1
$1\bar{1}01$	strong	1
$2\bar{2}00$	weak	0

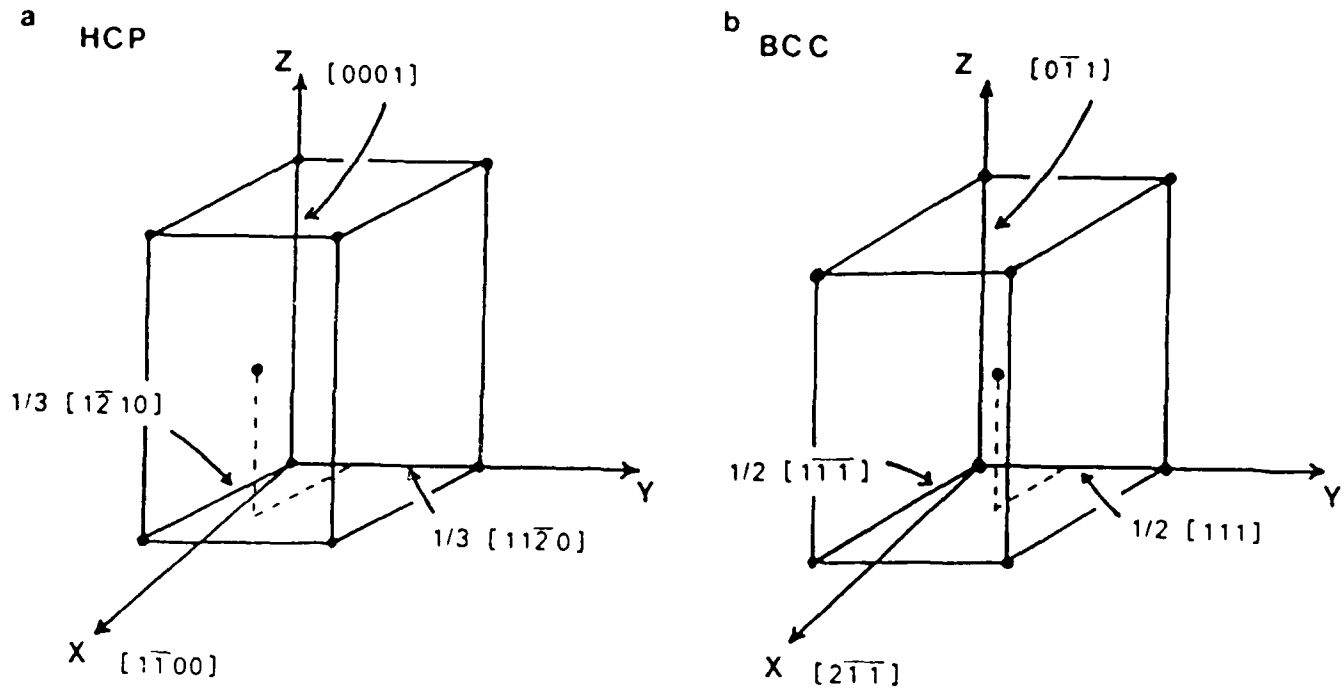


Figure A1: The unit cells chosen in the hcp and bcc lattices for the O-lattice calculations.

Table A1: The unit cells parameters used for the O-lattice calculations

Lattice parameter	Coordinates of hcp	Coordinates of bcc
a	$a_\alpha/3\,[1\bar{2}10]$	$a_\beta/2\,[1\bar{1}\bar{1}]$
b	$a_\alpha/3\,[11\bar{2}0]$	$a_\beta/2\,[111]$
c	$c_\alpha\,[0001]$	$a_\beta\,[0\bar{1}1]$
α	90°	90°
β	90°	90°
γ	109.47°	120°